Mechanism of the silicon influence on chilling tendency index and chill of ductile iron in thin wall castings

In the present work an analytical expression that combines the susceptibility of liquid cast iron to solidify according to the Fe-C-X metastable system (also known as the chilling tendency of cast iron), is proposed. A relationship between the chilling tendency index of cast iron and several factors has been presented. The results can be also used as a guide for a better understanding of the effect of technological variables such as the melt chemistry, the holding time and temperature, the spheroidizing and inoculation practice, the resulting nodule count and the type of mold material and pouring temperature, on the resultant chill of the ductile iron. Theory was experimentally verified using silicon as an example. In particular, it has been shown that as a result of increasing silicon content the critical nodule count increases, and the temperature range between the graphite eutectic equilibrium temperature and formation temperature for cementite eutectic, such variations lead to decreasing the chilling tendency index and in consequence reducing chills in cast iron. The chilling tendency index has been related to the critical wall thickness, below which the chill is formed. Theoretical calculations of the critical wall thickness were made and then compared with experimental outcome for ductile iron melts.

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1 Introduction

Significant efforts have been conducted recently towards the study of several aspects related to the production of thin wall ductile iron (TWDI) in order to introduce ductile iron pieces into the light parts market. As for all gray irons both, kinetic and thermodynamic factors influence the solidification structure and the final microstructure of TWDI. Very fast cooling rates, as those imposed during the solidification of TWDI parts, increase eutectic undercooling and may cause the formation of ledeburitic carbides, while slow cooling favors graphite precipitation. On the other hand, thermodynamic factors are related to the influence of alloying elements, particularly silicon. The susceptibility of the melt to solidify according to the Fe-C-X metastable system is known as the chilling tendency. In the foundry practice, the chilling tendency for the various types of cast irons is determined from comparisons of the exhibited fraction of cementite eutectic (chill) in castings solidified under similar cooling rate. Figure 1 gives a comparison of the chilling tendency for two cast irons (I and II). Cast iron I exhibits a lower chilling tendency than cast iron II. Based only on these comparisons, the difference in the chilling tendency of various cast irons can be established.

In particular, cast irons possessing a high chilling tendency are prone to develop zones of white or mottled iron. Considering that these regions can be extremely hard; their machinability and mechanical properties can be severely impaired. Alternatively, if white iron is the desired structure a relatively small chilling tendency favors the formation of grey iron which in turn leads to poor hardness and wear properties for the cast components. Hence, considerable efforts have been made in correlating various factors of technological relevance, such as chemical composition [1-4], pouring temperature [2], spheroidization and inoculation practice [2, 3, 5, 6], casting geometry [7], plate thickness [2, 3, 7], mold material [8], and nodule count with the chill of cast iron. These experimental relationships are very useful but they are limited in their physical meaning. Accordingly, in this work an analytical expressions is presented aiming to estimate quantitatively the chill formation. The main objective of this paper is to evaluate a new derived chilling tendency index of cast iron which can be theoretically calculated and related to of the critical wall thickness below which the chill is formed and verified by experiments.

2 Experimental procedure

TWDI plates, cast for a previous work [4] at the foundry pilot plant of the INTEMA Research Institute, have been used to validate the theoretical calculations. Three unalloyed ductile iron melts were produced by using a 55 kg capacity medium frequency induction furnace.

Charges were made using regular quality raw materials. The melts were superheated up to 1550 °C before tapping. Noduleization was carried out using the sandwich method and 1.5% of Fe-Si-Mg (6% Mg). They were inoculated with 0.6% Fe-Si (5% Si) in the stream. The melts had slight differences in carbon content and noticeable variations in silicon percentage, as is indicated in Table 1. The plates were cast using a horizontal model. The moulds were made of resin bonded 60/62 sand and the inner surfaces were coated with graphite paint. Six plates of 120 × 40 mm
were obtained from each mould. Three of them had 1.5 mm thickness, while the others had 2, 3 and 6 mm thickness, respectively. Figure 2 shows a casting which includes a fluidity spiral used to evaluate the castability.

As it can be seen all the plates, including the thinner ones were cast without significant cast defects. The microstructure of the samples was characterized on surfaces obtained by cutting the plates in the central zone. The microstructure characterization was aided by the use of the Image Pro Plus software. Nodule count was measured on unetched samples considering a nodule diameter threshold of 5 microns. In ductile iron the graphite nodules are characterized by Raleigh distributions [9] so the volumetric nodule count (nodules count per unit volume), \( N_v \), can be related to the planar nodule count (nodules count per unit area), \( N_p \) using the Wienczek equation [10]

\[
N_v = \frac{N_p}{f_g}
\]

where \( f_g \) is the volume of graphite at room temperature, \( f_g = 0.11 - 0.14 \).

The amount of carbides (area percentage) has been measured after etching with nital. Reported values of nodule count and carbide content are the average of at least five readings on each sample, at \( \times 100 \) magnification. As an example, Figure 3 shows the microstructure of a sample of 3 mm thickness corresponding to melt 1.

3 Analysis of results and discussion

By combining heat extraction and heat generated during solidification of spherical eutectic laws with kinetic growth laws and nodule count the solidification process of ductile iron has been calculated [11]. This analysis indicates that the critical wall thickness \( s_c \) below which the chill is formed, can be given by

\[
s_c = 2p \frac{CT}{\pi}\n\]

where

\[
p = a \left( \frac{T_i^3}{4 \pi^5 \beta B^2 L_e z^2 c^4} \right)^{1/6}
\]

\[
CT = \frac{1}{D^{1/2}} \left( \frac{1}{N_{\text{cr}} \beta} \Delta T_e \right)^{1/3}
\]

\[
\Delta T_e = T_e - T_c
\]

\[
B = \ln \left( \frac{T_e}{T_s} \right)
\]

\[
z = 0.41 + 0.93 B
\]

In the above equations: CT is the chilling tendency index of cast iron, \( T_i \) is the initial metal temperature just after filling the mould, \( N_{\text{cr}} \) is the volumet-

![Figure 1: Castings according to ASTM standard for chill and chilling tendency estimation](image)

![Figure 2: Casting showing the plates of different wall thicknesses](image)

<table>
<thead>
<tr>
<th>Table 1: Selected thermophysical data [11, 12]</th>
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<tbody>
<tr>
<td>Parameter</td>
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<tr>
<td>Latent heat of graphite eutectic</td>
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C, Si, P - content of carbon, silicon and phosphorus in cast iron, respectively, wt %
 Casting alloy, behaviour

Figure 3: As cast microstructure of a 3 mm thickness sample of melt 1, containing 9 % carbides (etched with nital 2 %)

ric critical nodule count at temperature $T_{m} = T_{c}$ (when chill appears in the critical wall thickness $s_{c}$, as shown in Figure 4b). D is the diffusion coefficient of carbon in austenite, and $\beta$ is the coefficient (see Equation (15)). The terms $T_{c}, c, a, T_{m}$, and $T_{c}$ are defined in Table 1.

Table 2 shows results, for both, chemical composition and wall thickness, as well as the exhibited nodule count and cementite fraction of the melts used.

3.1 Chilling tendency index CT

From our experimental data and the theoretical perspective, the role of the silicon on the chilling tendency index CT of ductile iron can be disclosed based on Equation (4).

3.1.1 Influence of the temperature range $\Delta T_{m} = T_{c} - T_{m}$

$\Delta T_{m}$ (Figure 4a) range depends on the melt chemistry (Table 1). For the melts used (Table 2) the values of carbon and phosphorus content range from 3.31 to 3.45 and from 0.044 to 0.051 %, respectively. Taking into account average values $C = 3.38$ % and $P = 0.047$ %, $\Delta T_{m}$ can be described by

$$\Delta T_{m} = 11.3 + 18.8 \text{ Si} \quad [\text{°C}]$$

(8)

It can be observed that as Si contents increase, the $\Delta T_{m}$ range also increases, and Equation (4) indicates that the chilling tendency index CT decreases.

3.1.2 Influence of the diffusion coefficient of carbon in austenite, D

This coefficient is related to solidification temperature of eutectic, $T_{m}$ and chemical composition of the austenite. The effect of Si, Mn and P on D is not considered in this work, as there is not enough information available in the literature.

An expression for the diffusion coefficient of carbon in austenite has been taken from the literature [13] which is given by:

$$D = \left(0.00453 + \frac{3.33957}{273.3 + T_{m}}\right) \exp\left(\frac{3.37065 - 15176.273}{273.3 + T_{m}}\right)$$

[cm²/s] (9)

where $T_{m}$ is the eutectic solidification temperature [°C].

The change of silicon content from 2.70 to 4.42 % (see Tables 1 and 2) can modify theoretically the eutectic solidification temperature, $T_{m}$ during the eutectic transformation from $T_{m} = 1176$ °C to $T_{m} = 1084$ °C. So, for these values of $T_{m}$ the D values in accordance with Equation (9) range from $2.7 \times 10^{6}$ to $5.6 \times 10^{6}$ cm²/s. Therefore, as Si content increases, D decreases, and from Equation (4) result that the chilling tendency index increases.

3.1.3 Influence of the critical nodule count

The critical nodule count is represented by $N_{c}$ in Figure 4b. It is well known that each graphite nucleus gives rise to a single nodule, so it can be assumed that the measure of

| Table 2: Chemical composition, wall thicknesses, nodule count, cementite fraction and chilling tendency index CT |
|---|---|---|---|---|---|---|---|---|
| Melt no | C, wt % | Si, wt % | P, wt % | Wall thickness, mm | Nodule count, NF, mm² | Fraction of cementite, % | Chilling tendency index, CT, $\delta^{10^3}$°C⁻¹ |
| | | | | experimental s | calculated $s_{c}$ | | |
| I | 3.40 | 2.70 | 0.046 | 6.0 | 588 | 0 | 0.68 |
| | | | | | 3.0 | 1039 | 9.0 | - |
| | | | | | 2.0 | 1380 | 24.0 | - |
| | | | | | 1.5 | 1311 | 34.0 | - |
| II | 3.45 | 2.91 | 0.044 | 6.0 | 854 | 0 | 0.60 |
| | | | | | 3.0 | 1037 | 7.3 | - |
| | | | | | 2.0 | 1100 | 24 | - |
| | | | | | 1.5 | 2027 | 9.3 | - |
| III | 3.31 | 4.42 | 0.051 | 6.0 | 1127 | 0 | - |
| | | | | | 3.0 | 1726 | 0 | - |
| | | | | | 2.0 | 1890 | 0 | 0.34 |
| | | | | | 1.5 | 2027 | 9.3 | - |

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graphite nuclei count, is the conventional nodule count. According with Table 2, in melt I the transition from a wall thickness, s = 6 mm (without cementite) down to 3 mm (with cementite) is closely linked to a nodule count change from 888 to 1039 mm⁻². As a result, an average nodule count value of $N_{Si}$ = 813 mm⁻² was used in this work. Similar determinations were made in melts II and III and $N_{Si}$ values of 945 and 1959 mm⁻² has been obtained, respectively. **Figure 5** shows the relation between silicon content in cast iron and the critical nodule count, $N_{Si}$. This relationship can be described by

$$N_{Si} = 655.9 \text{ Si} - 982.9 \quad [\text{mm}^{-2}] \quad \text{or}$$

$$N_{Si} = 10^3 (2.26 + 1.08 \text{ Si}) \quad [\text{cm}^{-2}] \quad (10)$$

$N_{Si}$ is the volumetric critical nodule count.

Taking into account Equations (4) and (10) it can be concluded that as silicon content increases the $N_{Si}$ also increases and in consequence the chilling tendency index decreases.

### 3.1.4 Influence of the coefficient, $\beta$

From theoretical analysis of spherical eutectic growth [11] result the following equation

$$k = \frac{(C_1 - C_2)}{(C_4 - C_3) \sqrt{\frac{C_2 - C_3}{C_4 - C_3} - 1}} \quad (11)$$

where $C_1$, $C_2$, $C_3$, and $C_4$ are carbon content defined in **Figure 6**.

The above expression can be correlated with the degree of undercooling, $\Delta T$. Assuming that the $\text{JE'}$, $\text{ES'}$, and $\text{BC'}$ lines for the Fe-C-Si system (Figure 6a) are straight compositions in Equation (11) can be given by

$$C_3 = C_{E'} - 0.11 \text{ Si} \frac{\Delta T}{m_2} \quad (12)$$

$$C_3 = C_{E'} - 0.11 \text{ Si} + \frac{\Delta T}{m_3} \quad (13)$$

$$C_4 = C_{C'} - 0.11 \text{ Si} - \frac{\Delta T}{m_4} \quad (14)$$

where: $m_2$, $m_3$, and $m_4$ are the slopes of the line $\text{JE'}$, $\text{ES'}$ and $\text{BC'}$ respectively and $C_{E'}$, $C_{C'}$ are the carbon content in austenite and eutectic for Fe-C system, Si is the silicon content (wt %).

The following values can be employed [14]: $C_{E'} = 4.26$ wt %, $C_{C'} = 2.08$ wt %, $m_2 = 275 [^\circ\text{C}$/wt %], $m_3 = 189.6 [^\circ\text{C}$/wt %], and $m_4 = 113.2 [^\circ\text{C}$/wt %]. Using this data, Equation (8) are plotted in **Figure 7a**. From this figure, it is apparent that $k$ tends to

**Figure 5**: Correlation between silicon content in cast iron and critical nodule count, $N_{Si}$. 

**Figure 4**: a) Cooling curves and b) effect of the wall thickness on the minimal eutectic solidification temperature, $T_m$ and nodule count $N$. $N_{Si}$ and $s_c$ are the critical nodule count and the critical wall thickness.

**Figure 6**: Theoretical analysis of spherical eutectic growth for Fe-C-Si system.
exhibit a linear trend with $\Delta T$. Accordingly, $k$ can be described by

$$k = \beta \Delta T$$

(15)

where $\beta$ are the slopes of the lines $k(\Delta T)$ in Figure 7a.

The effect of Si content on $\beta$ values is given in Figure 7b. However, it can be shown by calculations that the effect of Si on $s_c$ and $CT$ through $\beta$ in Equations (15) and (4) is very small. Thus, it can be assumed that $\beta$ is constant, $\beta = 0.00155\ ^\circ\text{C}^{-1}$. This assumption gives the error about $\pm 3\% \pm 6\%$ in our calculations.

All above parameters ($N_c, c_t, \Delta T_{ct}, D$) depend on silicon content. So, the chilling tendency index can be presented as silicon function. For the calculation effect of silicon content on chilling tendency index, $CT$, the values of $D$ ranging from $2.7 \times 10^6$ to $5.6 \times 10^6$ cm$^3$/s and Equations (4), (8) and (10) can be used. The results of these calculations are shown in Figure 8a. Thus, it can be stated that as silicon content increases the chilling tendency index of ductile iron decreases. It is quite well known that the chilling tendency determined by classic method of wedges decreases as silicon content increases. Thus, it can be stated that the present results reproduce the experimental knowledge.

### 3.2 Critical wall thickness $s_{cr}$

From Table 2, it is apparent that in melts I and II the chill occurs in walls with thicknesses between 3 and 6 mm.
while in melt III it happens at wall thicknesses between 1.5 and 2 mm. Hence, in order to compare these results with the theoretical predictions, estimations of $s_c$ were made, using Equation (2). In these calculations it was assumed that $D = 2.7 \times 10^{-6} - 5.6 \times 10^{-6}$ cm$^2$/s, $a = 0.11$ J/cm$^2$ s$^{1/2}$ °C and $T_i = 1250$ °C. Other relevant information was taken from Table 1. Results of these calculations are shown in Figure 8b. From this figure, it can be observed that as Si contents increase from 2.70 to 4.42%, the critical wall thickness $s_c$ decreases from 2.9-4.2 to 1.5-2.1 mm. In addition, a comparison calculated $s_c$ indicates that the predictions from the theoretical analysis are rather in good agreement with the experimental data.

4 Conclusions

A simple theoretical analysis which enables the prediction of the chilling tendency index and chill in ductile cast iron has been presented. Theory was experimentally verified using silicon as an example. In particular, it has been shown that as a result of increasing silicon content nodule count, $N_0$ and the critical nodule count, $N_c$, as well as the temperature range, $\Delta T_i$, increases. Such variations lead decreasing the chilling tendency index, CT and in consequence reducing chills in cast iron. The chilling tendency index, CT has been related to the critical wall thickness, $s_c$ below which the chill is formed. Theoretical calculations of $s_c$ were made and then compared with experimental outcome for ductile iron melts.

The predictions of the theoretical analysis are in rather good agreement with the experimental data. Presented model is general and a real challenge would have been to apply this approach to predict chill tendency index in Fe-Cr-Ni or Fe-C-Cu systems.

Figure 8: a) Influence of silicon content on chilling tendency index CT in cast iron and b) relation between silicon content and the critical wall thickness $s_c$ in ductile iron shadow area – calculated for $D$ ranging from $2.7 \times 10^{-6}$ to $5.6 \times 10^{-6}$ cm$^2$/s

Literature

Metal expansion penetration on concave casting surfaces of grey cast iron cylinder heads

Cylinder heads have an extremely complex shape with large areas of concave casting surfaces. The concave casting surfaces are often associated with metal expansion penetration problems or other surface defects, e.g. surface shrinkage. The defects cause high production costs due to component rejection and increased setting time. This report presents an investigation of the microstructure in grey cast iron close to the sand-metal interface affected by metal penetration in a complex shaped casting.

The dominant penetration defect observed in the cylinder heads was expansion penetration. Even pre-solidification penetration and sand crack defects were observed. The microstructure found in the non-penetrated areas is typical for solidification of grey cast iron in sand moulds.

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1 Introduction

Metal penetration is one of the most important factors affecting the surface finish of cast components in the iron foundry industry. Metal penetration often incurs extra fettling costs, and in severe cases leads to scrapping of castings. The extra fettling times and costs vary considerably from foundry to foundry, and with the severity of the defect and the type of casting produced.

The general definition of metal penetration accepted by the foundry industry, as proposed by Draper and Gaindhar [1] is the condition in which cast metal has entered into the pore spaces of the mould and core beyond the mid-point of the surface layer of sand grains.

During eutectic solidification, cast iron solidifying with an austenite-graphite eutectic structure has been demonstrated to expand significantly. The expanding metal exerts considerable pressure on the mould walls. If the mould is rigid, and if the metal cannot be pushed back in the riser or the gates because they have solidified, the pressure results in metal penetration, called expansion penetration, into the mould. This type of penetration was first identified by Levelink and Julien [2] called it exudation penetration. They indicated that the expansion occurring during eutectic solidification may result in exudation of eutectic in locations where a solidifying metal shell does not obstruct it. This is especially so at "hot spots", such as in the corners of "L", "T" and "Y" sections, where the metal is still liquid at the time expansion occurs due to eutectic solidification. The expansion penetration depends upon the metallurgical characteristics of the solidifying metals and alloys. Expansion penetration is most common when the CE (carbon equivalent) is high. This has been reported by Levelink and Julien [2] and confirmed in an earlier work by the authors [3].

In the work mentioned above an abnormal eutectic cell structure was found in the neighbourhood of penetrated areas. Two different populations of eutectic cell sizes were found to be present. Between a larger population of eutectic cells, a smaller population was identified which indicated that the different populations nucleated at different times during solidification. The mechanism of nucleation for the double eutectic cell population is not understood but it was characteristic for the penetrated areas.

The theory of exudation or metal expansion penetration proposed by Levelink and Julien was developed further. Dööszi e. a. [4] suggested a new description for the metal penetration mechanisms, considering the nucleation and growth of both the primary austenite grains and the eutectic cells. According to these descriptions there are two different metal expansion penetration mechanisms. The first type occurs before the columnar to equiaxed transition of the primary austenite grains, whereas the second type occurred after the columnar to equiaxed transition. A transition between the two mechanisms was also found. Typical for the first type of penetration mechanism was the exudation of a perfect eutectic phase from the interendritic area to the mould surface, while in the second type of penetration mechanism an anomalous phase from the eutectic cell border was pushed to the mould interface, simultaneously with the deformation of the cast surface. These observations were made both by Levelink and Julien and by Dööszi e. a. on cylindrical test cups with different shapes of the mould surface exposed to penetration.
The aim of this paper is to present the extended work on evaluation of the microstructures in a complex shaped cylinder head cast in grey cast iron with respect to metal expansion penetration and to compare the observations with those made by Diószegi e. a. on the metal mould interfaces of the experimental test cups. The investigation of the cylinder head casting was carried out as part of a research project in collaboration between the University of Jönköping, Division of Component Technology and the Skövde Foundry of Volvo Truck Component Corporation.

2 Examination of cylinder heads

The cylinder heads studied were prepared on a standard moulding line using green sand as moulding material. The cores containing quartz sand were bound by an organic binder using SO₂ gas as catalyst. The geometry of the cylinder head is considered to be a complex shaped part, as shown in Figure 1.

The moulds were poured from a 1.5 ton ladle, after inoculation in the ladle with a standard inoculant containing strontium. The chemical composition of the inoculant is shown in Table 1. The amount of the inoculant added was 0.15 wt %.

The melt was a grey iron with the chemical composition shown in Table 2. After blasting the cylinder heads were investigated by an ocular inspection. The specimens for study were cut out from the areas where metal penetration had occurred. Some of these specimens are shown in Figure 2. The penetration area is indicated by red circles.

The specimens were ground and polished in order to investigate the graphite morphology by optical microscopy. After a first inspection the specimens were etched in a picric acid-based reagent at 110 °C in order to investigate the primary austenite and the eutectic cells.

3 Discussion

The extremely complex shaped cylinder heads showed different types of penetration related defects. The colour etching technique used allows the defects to be classified with respect to the mechanisms of defect formation. The defects observed are discussed below in chronological order with respect to the mould filling and solidification of the different phases.

3.1 Pre-solidification penetration

A minor group of penetration defects were identified as metal particles which had penetrated between sand grains (Figure 3). The microstructures clearly reveal that the particles solidified with a coarse primary austenite phase surrounded by an austenite-graphite eutectic phase. This type of penetration is classified as pre-solidification penetration. The presence of all phases indicates that the metal droplet has the same chemical composition as the original melt, and was forced to penetrate between the sand grains before solidification started. This type of penetration is believed to result from the conservation of momentum at the end of mould filling.

3.2 Metal expansion penetration

The predominant number of penetration events observed in the cylinder heads were due to metal expansion penetration. The microstructure of particles which had penetrated between the sand grains showed an exclusively eutectic composition (Figure 3b). The casting surfaces in connection to the penetrated particles with the eutectic phase consisted of a eutectic strip with graphite lamellas.

![Figure 1: A sectioned cylinder head](image)

![Figure 2: The investigated specimens showing metal expansion penetration.](image)
Casting defects

Figure 3: Microstructure of the penetrated area

Figure 4: Microstructure of the first type expansion penetration

Figure 5: Microstructure at the metal-mould interface with sparse fraction primary phase austenite and an anomalous interdendritic phase

Figure 6: Microstructure behind the abnormal structure presented in Figure 5. The fraction primary austenite is in good correlation to the chemical composition, and the graphite morphology is type A.
oriented mainly perpendicularly to the mould surface (Figure 4).

This combination of microstructure in the penetrated particles and the casting surface is identical to that observed by Diószegi e. a. [4]. Metal penetration was found to occur during the early part of the solidification process, before the columnar to equiaxed transition, and before the eutectic phase is squeezed through the primary austenite network of the outer columnar austenite grains to the mould surface. 

Behind the penetrated particles of eutectic composition there was observed a second type of microstructure in connection to the mould surfaces. This type of microstructure indicates a normal fraction of primary phase behind the casting surface (Figure 5) and much less dense primary phase in connection to the casting surface (Figure 6). The normal fraction of primary phases is determined by the chemical composition.

### 3.3 Cracks in the core

The microstructure investigations of the metal expansion penetration close to the sand cracks (Figure 7) show two different types of microstructure. The first type of microstructure is found when molten metal filled the crack in the core, and consists of white solidified iron. Figure 8a. This type of crack in core is assumed to happen very early during the mould filling when thermal shocks are believed to contribute to the sand crack formation. The completely white solidification of the metal in the crack indicates a very quick solidification.

The second type of microstructure occurs when core cracks are connected with metal penetration between the sand grains, and is shown in Figure 8b. The metallic particles between the sand grains and the bridge connecting the penetration area to the casting surface contain exclusively a eutectic phase. This indicates that the moment of penetration is at a late stage of solidification, when only the eutectic phase can exist. In contrast to the case when thermal shock leads to distortion of the core surface, in the latter case the core is already heated, and it is probable that the grain boundaries are weakened and crack formation is caused by distortion of the outer austenite shell.

### 3.4 Double eutectic population

A double population of eutectic cells was found (as shown in Figure 9) close to the penetrated areas, but in some cases it also occurred in non-penetrated areas. It is difficult to clearly understand the relationship of this phenomenon with penetration in these castings.

Figure 7: The specimens showing cracks in the core

Figure 8: Microstructure of the penetrated area
3.5 Non-penetrated areas vs. surface shrinkage

Convex core surfaces associated with penetration problems are sometimes also affected by surface shrinkage. Figure 10 shows examples of surface shrinkage found in the cylinder head. The microstructure behind the surface shrinkage shown in Figure 11a is compared to the microstructure of concave metal surfaces where penetration was not found, in Figure 11b. It is difficult to see any differences. Both microstructures appear to be normal with respect to similar casting structures of comparable chemical composition and cooling conditions.

4 Conclusions

Cylinder heads typically have an extremely complex shape, with large areas of concave casting surfaces which are prone to surface defects. Investigations made on cylinder heads cast under industrial production conditions reveal the same modes of expansion penetration as have been found to occur in experimental test cups cast at the same foundry and under the same metallurgical conditions. Metal expansion penetration which is deduced to occur before and after the columnar to equiaxed transition is observed to predominate. Presolidification penetration and sand crack defects were also observed. The microstructures below the surface shrinkage appear to be identical with microstructures of concave casting surfaces not affected by penetration.

The frequency of penetration defects observed shows the expansion penetration as being the dominant surface defect. Only a minority of penetration defects found can be traced to phenomena prior the solidification event.

For measures to prevent penetration defects the authors would like to recommend close metallurgical control of the melting process, including adequate control of nucleation and solidification.

The authors would like to thank the partner of this research project, namely the Skövde Foundry of Volvo Truck Component Corporation.

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Figure 11: a) Colour etched microstructure close to the shrinkage defect; b) microstructure of the non-penetrated area

Literature